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TECHNICAL NOTE 3728

STUDY OF ALUMINUM DEFORMATION BY ELECTRON MICROSCOPY

By A. P. Young, C. W. Melton, and C. M. Schwartz

Battelle Memorial Institute

Washington
August 1956

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SUMMARY

In this investigation, the slip-line structure in a series of aluminum single crystals deformed in tension was examined in the electron microscope. In several of the crystals the slip-line structure was examined after successive steps in the elongation.

In formulating a mechanism for slip-band formation, the experimental results of this study and of other investigations were considered. According to the proposed mechanism, slip in aluminum single crystals originates from a relatively few Frank-Read sources present before deformation in the annealed crystal. Dislocation loops are generated by these sources on a few planes of the set of parallel slip planes with the highest resolved shear stress. Frank-Read sources for slip on neighboring parallel planes are formed by the deflection of screw segments into oblique planes in a manner suggested by previous workers. The geometry of this deflection mechanism is discussed in some detail to show that it may be feasible theoretically, even when the dislocations are extended as they probably are in aluminum. The basic difference in the slip-line structure of aluminum or copper and alpha brass, as disclosed by previous workers, is accounted for by showing that, in the latter case, short-range order may prevent the deflection of screw segments. Other differences between face-centered cubic pure metals and their substitutional alloys are discussed.

INTRODUCTION

This investigation had several objectives:

(1) Development of a nondestructive replica technique for electron-microscope examination of slip-line structure

(2) Determination of the slip-line structure in deformed aluminum

(3) Development of a theory of plastic deformation in aluminum

The development of a nondestructive replica technique for electron microscopy is important because it makes possible the examination of a crystal after various degrees of deformation. A nondestructive technique was developed in this investigation, and it was used to follow the development of slip-line structure in aluminum crystals. Ways in which this technique can be used to greater advantage than it was used in this investigation are suggested in the report. This replica technique has another advantage over previous techniques, which depended on a chemical separation of the replica from the metal crystal. For this chemical separation, a chemical had to be found which would attack the metal but not the replicating film, which was either metal oxide or silica. This is not possible for noble metals such as gold or platinum and probably would not be possible for titanium, zirconium, and many other metals.

There has been considerable disagreement among various investigators about the nature of the slip-line structure in deformed aluminum crystals. There is good evidence that mechanical polishing may affect the slip-line structure (ref. 1). The crystals used in this investigation were not mechanically polished after annealing but were electropolished in an acetic-perchloric bath.

In the development of a theory of plastic deformation the data from this investigation were considered in combination with X-ray and other data of various investigators.

This investigation was conducted at Battelle Memorial Institute under the sponsorship and with the financial assistance of the National Advisory Committee for Aeronautics. The authors wish to thank Mr. R. D. Smith for his part in the experimental work and Mr. L. L. Marsh for furnishing the strain-annealed crystals for the investigation. Thanks are also extended to Professor B. Chalmers and Dr. R. J. Harrison for helpful discussions and suggestions during the course of the work.

MATERIALS AND TEST SPECIMENS

The material used in this investigation was aluminum of 99.99-percent purity. Single crystals grown both by the strain-anneal method and from the melt were used. The strain-annealed crystals had a round cross section 0.505 inch in diameter. The method of growing these crystals has been previously described (ref. 2). The crystals grown from the melt had a square cross section, approximately 1/4 by 1/4 inch. These crystals were obtained from Mr. M. Lauriente of The Johns Hopkins University. They were grown in a mold by the Bridgman method (ref. 3) and were single crystals over a gage length of approximately $2\frac{1}{2}$ inches. The crystals

denoted by a letter plus a number are the strain-annealed crystals (e.g., crystals S68 and P64). The other crystals are denoted by a number only (e.g., crystal 4).

EXPERIMENTAL METHODS

The aluminum crystals to be examined were electropolished with no previous mechanical polishing. The crystals were elongated in a soft machine by dead-weight loading. After elongation, replicas for electron microscopy were made. Some of the crystals were then repolished and elongated again to reveal slip-line structure in prestrained crystals. Replicas were made at each stage of elongation.

The electropolishing solution was the same as the one used by the Wilsdorfs (ref. 4) (30 cubic centimeters of perchloric acid and 100 cubic centimeters of glacial acetic acid). Polishing was done at 10° C and 13 volts.

Faxfilm[1] impressions of the surface were used for both light and electron microscopy. Silica-backed metal replicas for electron microscopy were made by vacuum-evaporating heavy metal on the Faxfilm impression at a 1-to-1 shadow angle. A thin backing film of silica was then evaporated at normal incidence. The purpose of the silica was to increase the strength and rigidity of the replicas. The Faxfilm was dissolved in acetone by a method essentially the same as that described elsewhere for plastic replicas (ref. 5). After solution of the plastic, the composite metal-silica replicas were ready for viewing in the electron microscope. Judging from the fineness of some of the visible slip-line structure, there is no appreciable loss of detail in the plastic impressions. The granular appearance in some of the pictures is due to the shadowing metal. Gold or palladium is easy to evaporate, but in each case the film is granular (ref. 6). A platinum film is not granular, but it is difficult to evaporate. Simultaneous evaporation of various combinations of these metals was tried. The film was granular in some instances. A combination of 75 percent platinum and 25 percent palladium seemed to offer the best compromise between reasonable ease of evaporation and lack of granularity.

For light microscopy, gold was evaporated on Faxfilm impressions. Pictures were taken by transmitted light. Large-scale undulations in the electropolished surface and variations in thickness of the plastic film caused some of the light micrographs to appear blotchy. However, the fine detail is not obscured.

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[1]A plastic tape obtainable from the Brush Electronics Co. To replicate, the tape is moistened on one side with acetone and pressed against the surface with the fingers until it sets.

The replicas for electron microscopy were supported on a fine wire mesh. They were examined under the light microscope to correlate the spacing of slip bands with the size of the opening in the wire mesh. Only replicas which showed the same spacing of bands in the electron microscope were used. This precaution was to assure that the fine structure seen in the electron microscope was really representative of the overall structure.

RESULTS AND INTERPRETATION

In the discussion of the results the terms "slip band" and "slip line" have particular meanings. The term "slip band" denotes the marking which appears as a single line in the light microscope. It is now known that this line marking is resolvable in the electron microscope into a band of slip lines which are apparently traces of slip on single planes. The terms "primary slip plane" and "oblique plane" are used in the discussion. In a single crystal, which is not oriented for duplex slip, slip at moderate deformation is primarily in one direction and on a parallel set of (111) planes. A plane on which slip takes place is called a primary slip plane. An oblique plane is a (111) plane which intersects a primary slip plane in a line parallel to the slip direction.

The Wilsdorfs (ref. 4) discovered that slip lines in deformed aluminum appear not only in the slip bands but almost all over the surface of the crystal. They called the fine slip outside of the bands elementary slip. Figures 1 and 2 of aluminum crystals deformed at room temperature verify that there is fine slip outside of the bands. However, the Wilsdorfs' statement that the fine slip is nearly uniform over the surface is not verified. There actually appear to be zones of fine slip. These are particularly noticeable in figures 2(a) and 2(b). The distinction then between slip inside and outside of bands is rather blurred.

At 1.5-percent elongation there is very little indication of slip bands in the light microscope. However, examination of figure 1(a) indicates that slip is already somewhat localized. There are, however, wide gaps between all the slip lines. Apparently, some of these gaps begin to fill in with slip lines to form bands at higher stress, as shown in figures 1(b) and 2(a). This interpretation is not based on examination of the same area after successive elongations and admittedly is somewhat speculative.

If it is assumed that step heights on the surface less than 20 angstroms high would not be detectable in the electron microscope, then the visible slip lines represent slip of one plane over the next of about 10 slip vectors. The assumption is made here that slip of this magnitude takes place by the generation of dislocation loops from a Frank-Read

source (ref. 7, pp. 69-90). It has been suggested by Koehler (ref. 8) and Orowan (ref. 9) that an entire slip band can be generated from one Frank-Read source originally in the crystal before deformation. A screw segment of a dislocation can slide off a primary slip plane into another primary slip plane via an oblique plane. The process is illustrated in figure 3. The deflection of the screw segment into another plane in the interior of the crystal will not result in any surface evidence of cross-slip.

In a face-centered cubic metal this deflection process will be inhibited because the dislocations are probably extended and will tend to stay in the primary slip plane. However, when moving dislocations intersect screw dislocations with a different Burgers vector, segments of the moving dislocations are transferred to the next parallel plane. These jogs (see, e.g., fig. 12(b)) retard the motion of screw dislocations in the slip plane because the jogs, which are in the edge orientation, leave behind trails of vacancies or interstitials (ref. 7, pp. 69-90). That screw dislocations probably do move more slowly than edge dislocations has been shown by Chen and Pond (ref. 10) and by Becker (ref. 11) in their investigations of the dynamic formation of slip bands in aluminum. The slowing up of the dislocations with screw components makes it likely that expanding loops from a Frank-Read source are oblong. Fairly long screw segments may be formed. The outside loop will be the most retarded in its motion and the other loops will pile up behind it. When a long screw segment is nearly stopped in a primary slip plane, it may be deflected as shown in figure 3.

The jogs not only stop the progress of the screw dislocation in the primary slip plane but also provide the mechanism by which the extended dislocation can be deflected. The mechanism by which screw segments may be deflected is described in more detail in appendix A.

In figures 1 and 2, the spacing between slip lines is obviously not uniform in either the fine or the coarse slip regions. However, there are what might be termed subbands of slip of similar spacing and step height. This is particularly noticeable in figure 1(b) and in figure 2.

Deflection of a screw segment into an oblique plane will not insure slip in a parallel primary slip plane. A screw segment in the oblique plane must in turn be formed and move off the oblique plane into a primary slip plane. This may take place fairly readily because the primary slip planes are the planes of high resolved shear stress. The stress required for the screw segment to be activated as a Frank-Read source is inversely proportional to the length of the source (ref. 7, pp. 69-90). It is expected that at low stress the loop in the oblique plane may be quite large before a segment long enough to act as a Frank-Read source is formed. The first slip planes in a band may then be quite widely

separated, as shown in figure 1(a). The band will fill in with more closely spaced slip planes as the stress increases. In order to verify this hypothesis, it is evident that what is needed is a step-by-step investigation in the electron microscope of the formation of individual slip bands. This would be a tedious job, but it is entirely feasible with the nondestructive replica technique used in this investigation.

The Frank-Read sources formed by deflection are in the screw orientation. It is very likely, then, that after the generation of a number of dislocation loops the source itself will be deflected into another primary slip plane. If this happens a few times, some regularity in the spacing and step heights of the slip lines might be expected. This may explain the subbands of regularly spaced slip lines observable in so many of the micrographs.

Examination of the slip-line structure at 1.5- and 4-percent elongation indicates that the formation of slip bands may already be quite extensive. Figure 4(a) is a light micrograph of crystal P68 with 5-percent elongation. A high density of slip bands with a range of width is visible in this photograph. The question arises whether this evidence of extensive slip-band formation is a surface phenomenon or is indicative of the condition in the interior of the crystal. If Frank-Read sources generate bands throughout the crystal, there will be islands of slip. On leading and trailing edges of these islands, edge dislocations of opposite sign are piled up. These pileups result in local curvature in the slip planes and tilting of blocks of the crystal between pileups of opposite sign. The local curvature and tilted blocks could account for the X-ray asterism found by Honeycombe (ref. 12) in an aluminum crystal deformed in tension. Honeycombe noticed some blurring of Laue reflections after 1-percent extension and very marked asterism after 4- to 5-percent extension. Apparently, even for small extensions, slip takes place in the interior in bands and not by uniform fine slip as postulated by the Wilsdorfs (ref. 13).

Figure 4(b) is a light micrograph of an aluminum crystal elongated 20 percent. Comparison of figures 4(a) and 4(b) indicates that the slip-band density may not be much greater at 20-percent elongation than it is at 5-percent elongation. Examination of several samples in the electron microscope indicated that under increasing stress the slip bands widen with the addition of more slip lines and that some of the already existing slip lines grow deeper. Without following the progress of slip lines in the same area, the exact manner of slip-band formation must be somewhat a matter of conjecture.

Whether or not most of the slip bands are nucleated at fairly low strain depends on the initial distribution of Frank-Read sources in the crystal.

Not much is known about the distribution of dislocations in an annealed metal crystal. The most reasonable assumption is that the dislocations in a cubic lattice form a three-dimensional network. Visual evidence of such a network has been found in a silver bromide crystal (ref. 14). The resolved shear stress needed to activate a Frank-Read source is inversely proportional to the length of the source. The resolved shear stress on the primary slip planes at 5-percent elongation is approximately five times the critical resolved shear stress. Some appreciable fraction of the Frank-Read sources must be activated at or near the critical resolved shear stress. If the sources are segments in a three-dimensional network, there are probably not many with lengths less than one-fifth the length of the longest segments. It seems likely then that at 5-percent elongation most of the sources originally in the annealed crystal have been activated.

Becker (ref. 11) followed the growth of slip bands as a function of time by measuring the light intensity of the resulting lines. He found that with a constantly increasing stress the slip bands did not develop uniformly with time but increased in steps. This is in accord with the premise that the screw segment which is deflected may be too short to be an immediately active Frank-Read source. When the proper stress is reached, a few loops will be quickly generated. Becker's method was probably not sensitive enough to record the addition of a single slip line. It is more likely that he recorded the formation of a group of slip lines at approximately the same stress. This might occur when the Frank-Read sources in the screw orientation are deflected.

The simplest explanation of the fine slip-line structure in the deformed aluminum crystals is that it is a result of dislocation loops from Frank-Read sources which were originally farther below the surface than the sources which generate the stronger lines. This probably accounts for most of the fine lines. Some of them may be accounted for by another mechanism. Some of the screw segments which slide off the dislocation loops are probably too short to be activated as Frank-Read sources by the external applied stress. However, the applied stress may be reinforced by the stress from an adjacent pileup of screw dislocations. The force on a screw segment 1 micron from the center of a pileup of screw dislocations representing a total slip of 5,000 angstroms is approximately equal to the force exerted on the segment by the externally applied stress at 5-percent elongation.

This theory supposes then that all the slip in the crystal is generated from relatively few Frank-Read sources originally present in the crystal. The Wilsdorfs' theory (ref. 13) assumes that sources responsible for the fine slip are already present in the annealed crystal. This presupposes a dislocation density of about 10^8 dislocation lines per square centimeter (ref. 13). Recent X-ray data (ref. 15) indicate that

the dislocation density in an annealed crystal may be much lower than was previously assumed. The density may be as low as 5×10^6 lines per square centimeter. Admittedly, the interpretation of the X-ray data is questionable when so little is known about the actual distribution of dislocations in an annealed crystal. Some information, from X-ray diffraction, about the relative increase in dislocation density when an annealed crystal is slightly deformed in tension would be helpful in formulating a theory of deformation.

Strain-hardening in a deformed face-centered cubic crystal can arise from a combination of causes. Expansion of dislocation loops is retarded by the formation of jogs (ref. 7, pp. 69-90). Also, when the dislocation density is high enough, interaction between edge dislocations of opposite sign is likely. The stress required to move edge dislocations past each other on neighboring planes is dependent on the distance between the planes (ref. 7, p. 128). As the density of edge dislocations increases, greater stress is required to move them through the crystal.

By careful handling of the crystals and avoidance of mechanical polishing before extension, Rosi (ref. 16) has shown that in aluminum, copper, and silver crystals, not oriented for duplex slip, the stress-strain curve has the general character shown in figure 5. The curves show a region of easy glide up to between 2- and 4-percent strain, depending on the orientation of the crystals. This region of easy glide would be likely if most of the Frank-Read sources originally in the crystal were activated at low strain. The sources can generate a few loops at low stress before hardening by the formation of jogs takes place. Apparently, from figure 1, some of the longer segments which have been deflected are also activated at low strains. As the Frank-Read sources and the other longer segments are used up, increasing stress is required to expand the existing loops and to activate the shorter sources. The region of easy glide gives way to a region of more rapid strain-hardening.

The region of easy glide can be construed as evidence for the theory that nearly all the Frank-Read sources originally in the annealed crystal are used up in the early stages of deformation. It must be admitted, however, that many investigators have not observed this region of easy glide in the stress-strain curves of aluminum or copper single crystals. It is not apparent from the literature why there is such grave disagreement among observers regarding the shape of the stress-strain curves.

No mention has been made as yet of the relationship between kink bands and strain-hardening. No attempt was made in this investigation to examine the structure in a kink band in the electron microscope. Kink bands are probably nucleated by the tilt of a particularly large block between the leading and trailing edges of two islands of slip (ref. 17). Under certain conditions the edge dislocations may not move past each

other but are stuck in the crystal and become traps for other edge dislocations. Whatever the mechanism of kink-band formation may be, Nabarro (ref. 18) has remarked that kink bands are probably not very important in the strain-hardening mechanism because they are so far apart.

Electron micrographs were taken of aluminum single crystals elongated 20 percent. Some of these crystals have bands with 50 or more slip lines, as shown in figure 6. The electron-microscope examination confirmed the conclusion from the light micrographs that there is no marked increase in the number of bands between 5- and 20-percent elongation. The bands are much wider and apparently have filled in with slip lines. According to figures 6 and 7(a), there is no real basis for a theory of slip-band formation based on regularity of slip-line spacing and step height.

Figures 7(b) and 8 show the structure in a crystal which was elongated in steps. After each step, replicas were made and the crystal was repolished. Figure 9 shows the structure in a crystal strained 20 percent after a 20-percent tensile prestrain. There is an ambiguity in the interpretation of figures 8 and 9. The slip lines could be either from the expansion of loops previously formed in the crystal or from loops newly formed after the prestrain. One can surmise that the loops in the center of the band in figure 8(a) had formed and expanded before prestrain and that the loops on either side of the center were formed after prestrain and gave rise to a larger amount of slip than did the loops previously formed and expanded. A similar and more extreme case of the same condition may be true in figure 8(b).

These photographs do show that the slip still takes place in bands, even at high strain. It might be possible with the aid of locating marks to take electron micrographs of certain areas and then, after light repolishing and restraining, to take more pictures of the same areas. This should give more information about the development of slip bands.

Figure 10 shows the slip-line structure superimposed on the subgrain structure obtained when an aluminum crystal was chemically polished in Alcoa R5 Bright Dip.[2] There is no apparent effect of subgrain boundaries on the slip. The slip lines were not so well defined with this polish. There is probably a chemical film of some sort on the surface.

An attempt was made to deform aluminum crystals in both rapid strain and creep at 500° F. The samples were coated with silicone oil. Neither sample showed any resolvable slip-line structure in the electron microscope. The surface of the crystal was evidently oxidized. A crystal was then coated with silicone oil and strained rapidly at 500° F in a helium

[2]A patented chemical polishing formula licensed by Aluminum Company of America.

atmosphere. Some structure was visible in the electron microscope, but most of the bands showed evidence of a cracked oxide film. Finally, a crystal was rapidly strained at 250° F in a helium atmosphere. Figures 11(a) and 11(b) are electron micrographs made from this crystal. There is no really significant difference between the structure in the crystals deformed at room temperature and at 250° F. The bands in the crystal deformed at 250° F may be wider, on the average. Koehler (ref. 8) has proposed that the deflection of screw segments may be due in part to thermal fluctuations. This could account for wider slip bands at elevated temperature.

Some crystals should be deformed in both rapid strain and creep at a temperature higher than 250° F. Evidently, aluminum crystals must be strained in a high vacuum. It would probably be easier to use gold crystals for electron-microscope examination of face-centered cubic crystals strained at elevated temperatures.

Aluminum crystals strained in tension only a few percent exhibit considerable X-ray asterism. Cadmium crystals, on the other hand, exhibit very little asterism after nearly 100-percent elongation (ref. 12). Cadmium crystals strained in tension do not form slip bands as does aluminum. Sometimes there is an extraordinary amount of slip on single planes (ref. 19, fig. 3 opposite p. 8). This indicates that X-ray asterism is related to slip bands in the manner previously discussed. The absence of X-ray asterism and slip bands in cadmium can be explained by the fact that cadmium slips on basal planes. There are no oblique planes for the transfer of slip between primary slip planes by the deflection mechanism.

It has been shown that alpha-brass crystals strained in tension exhibit almost no X-ray asterism so long as slip is confined to a single set of planes (ref. 20). Also, the Wilsdorfs (ref. 21) have shown that the slip lines in alpha brass as seen in the electron microscope do not appear to be in bands as they are in aluminum. The slip appears to be for the most part on randomly spaced single planes and varies in step height from a few angstroms to a few thousand angstroms. The Wilsdorfs observed almost no slip in brass comparable with what they called elementary slip in aluminum and copper.

Alpha brass has the face-centered cubic structure. Therefore, oblique planes for the transfer of slip between primary slip planes are present. The deflection mechanism for transfer of slip between primary slip planes, which produces X-ray asterism and slip bands in aluminum, is apparently not operative in brass. This can be explained by a theory proposed by Fisher (ref. 22) and discussed more fully by Cottrell (ref. 23). Cottrell has explained why, in a crystal with short-range ordering, the initial motion of a dislocation in a slip plane requires a high resolved shear stress. Slip becomes easier as the dislocation moves through the crystal until, when the dislocation has moved a distance

one-half the average linear dimension of an ordered domain, slip is no longer blocked by short-range order. Briefly, this phenomenon is due to the fact that, when the dislocation advances through the first, third, and succeeding odd-numbered slip vectors, like atoms are moved into neighboring positions, thus raising the energy of the dislocation line. The effect decreases as the dislocation moves until, when the dislocation has moved half the average length of an ordered domain, as many like atoms are moved apart as are moved together.

In an alloy with short-range order, deflection of a screw segment into an oblique plane is unlikely because of the high energy barrier for the first step. The fact that there is neither slip-band structure nor fine line structure in brass is added evidence for the theory that both the slip-band structure and the fine slip in aluminum are probably products of the deflection mechanism which is operative in aluminum and not in brass.

There is X-ray evidence for short-range order in alpha brass (ref. 4). Also, the Wilsdorfs (ref. 21) have noted the similarity between the stress-strain curves of alpha brass and copper-gold alloys that are known to be ordered. There is also thermodynamic evidence for short-range ordering in brass (ref. 25).

Koehler (ref. 26) has noted that, when slip is initiated on one set of parallel planes in a brass crystal, slip will continue on the same set of planes until the crystal has rotated to the place where the resolved shear stress on a second set of planes exceeds the resolved shear stress on the first set by the amount of the yield stress. When slip does begin on a second set of planes, it proceeds much the same as it did on the first set. Prominent multiple slip lines have been seen by several observers in deformed brass in the light microscope. The Wilsdorfs (ref. 21) observed, in the electron microscope, multiple slip in brass in which the primary and secondary slip planes were indistinguishable.

This phenomenon is uncommon in aluminum or copper. The reason for this difference probably is that slip in aluminum and copper in the primary slip planes takes place by the generation of a great many dislocation loops of relatively small area. In brass, on the other hand, slip apparently occurs by the generation of relatively few loops of large area. For a given amount of slip, the density of screw dislocation in the primary slip planes in aluminum and copper is high, while it is relatively low in brass. The screw dislocations block slip on other planes in which both the slip direction and slip plane are different.

The deformation modes of copper and brass are then quite different. This is probably the explanation for the observed difference in rolling texture of copper and alpha brass (ref. 27). On the other hand, the

rolling texture of a copper-nickel alloy is nearly the same as the copper texture (ref. 28). The slip-line structure of the copper-nickel alloy should be examined in the electron microscope to see whether it is more nearly like the copper or the brass structure.

The deformation mode of alpha brass may explain its extreme ductility. In a polycrystalline material, slip in a grain has to take place on several slip systems if the grains are to maintain contact at the grain boundaries. Slip on any of the potential systems is probably relatively easy in brass because they are not hardened against multiple slip.

Several theories of solution-hardening have been proposed (ref. 29). Apparently, an alloy in which short-range-order hardening is predominant may have quite a different deformation mode from an alloy in which some other type of solution-hardening predominates. The difference in deformation modes will probably mean a difference in mechanical properties. It appears then that an investigation of slip-line structure and of the different types of solution-hardening in alloys may be of more than academic interest to the metallurgist.

Battelle Memorial Institute,
Columbus, Ohio, August 1, 1955.

APPENDIX A

MECHANISM OF DEFLECTION OF AN EXTENDED DISLOCATION INTO AN OBLIQUE PLANE

The energy of a straight dislocation line is increased by jogs because the extended dislocations are compressed at the nodes. Stroh (ref. 30) has suggested that the energy of a dislocation line can be lowered if jogs of unlike sign move together and annihilate each other while jogs of like sign join to make deeper jogs. Between two jogs of opposite sign a screw segment is already deflected into the oblique plane as shown in figures 12(a) and 12(b). If the segment is long enough, the force of attraction between the unlike segments may be counteracted by the tendency for expansion of the dislocation loop in the oblique plane. The segment between the jogs may be the total length of the deflected segment as in figure 12(a); or possibly a small section between deep jogs in a long screw segment as in figure 12(b) may expand and pull the remainder of the segment into the oblique plane. It can be shown that this latter possibility is unlikely.

Figure 12(c) is a magnified view of the jogs and intervening segment in figure 12(b). The crosshatching represents an extended dislocation in the horizontal slip plane. The partial dislocations have to come together at the jog. Before any more of the screw segment can slip off into the oblique plane, the partials have to be moved together.

In expanding the area of slip in the oblique plane, the dislocation will move and will be curved at the nodes as shown in figure 12(d). If it is assumed that the curved dislocation is not extended (ref. 7, p. 105), then its line tension T is approximately equal to the energy per unit length of a perfect dislocation:

$$T \approx Gb^2 \qquad \text{(A1)}$$

where G is the average shear modulus of the material and b is the magnitude of the slip vector (ref. 31). In order for any more of the screw segment to be deflected into the oblique plane the partial dislocations must be moved together. The sum of the unit-length energies of the two partials is about 2/3 T. The energy per unit length required to move the partial dislocations together is 1/3 T. This is the force opposing the motion of points A or B along the screw dislocation. The force at point B is $T \cos \theta$ where θ is the angle between the curved dislocation in the oblique plate and the undeflected screw dislocation is shown in figure 12(d). For $\cos \theta \geqq 1/3$ or $\theta \leqq 70^{\circ}$, an infinitesimal segment of the extended dislocation will close and move into the oblique plane.

If the farthest point of penetration of the dislocation loop into the oblique plane in figure 12(d) is at the distance d from the intersection of the planes, then for $\theta = 70^\circ$ the radius of curvature R at the node is approximately

$$R \approx \frac{d}{1 + \cos\theta} = 0.75d \tag{A2}$$

The force on the dislocation is $F = T/R$. Therefore,

$$F \approx 1.34\frac{T}{d} = 1.34\frac{Gb^2}{d} \tag{A3}$$

If τ is the resolved shear stress on the oblique plane, then

$$F = \tau b \tag{A4}$$

The shear stress required for $\theta = 70^\circ$ is

$$\tau = \frac{1.33Gb}{d} \tag{A5}$$

If τ were equal to the resolved shear stress on the primary slip planes at about 5-percent elongation, d would have to be about 4 microns. This is many times greater than the spacing between the slip lines in a slip band. It is unlikely then that deflection takes place by a short segment between jogs pulling the rest of the screw dislocation into the oblique plane.

If a long segment between jogs is deflected as in figure 12(a), then it is likely that a segment from an inner loop which has not crossed very many screw dislocations will be deflected. The sources themselves in the screw orientation will probably be deflected after enough loops are generated to create sufficient back stress.

If a long screw segment, as in figure 12(a), is deflected, the repulsive force between screw dislocations of the same sign will push the segment further into the oblique plane. This force will be in addition to the force due to the resolved shear stress in the oblique plane. The approximate magnitude of the repulsive force tending to push the screw segment into the oblique plane can be calculated as follows:

If n dislocations are piled up against an obstacle, the number of dislocations that can be packed into a length L of the slip plane is approximately (ref. 19, pp. 104-107)

$$n \approx \pi L\sigma/Gb \tag{A6}$$

where G is the rigidity modulus and σ is the resolved shear stress on the slip plane.

In a coordinate system where the xz plane is a primary slip plane and a screw dislocation is parallel to the slip axis, the stress components from the screw dislocations are

$$\tau_{xz} = -\frac{Gb}{2\pi}\frac{y}{x^2 + y^2} \tag{A7}$$

$$\tau_{yz} = \frac{Gb}{2\pi}\frac{x}{x^2 + y^2} \tag{A8}$$

If a deflected dislocation parallel to the z-axis has moved a distance D into the oblique plane, then

$$D^2 = X^2 + Y^2 \tag{A9}$$

where X and Y are the components of D along the x- and y-axes as shown in figure 13.

The undeflected stress dislocations shown in figure 13 represent segments and extend a distance L into a primary slip plane, the xz plane, where L is parallel to the x-axis. The stress at the deflected dislocation position is the sum of all the stresses from the dislocations in the primary slip plane. If the sum is replaced by an integral, then the stress components at the deflected dislocation position are

$$\tau_{xz} = -\frac{Gnb}{2\pi L}\int_X^{X+L}\frac{Y}{x^2 + Y^2}\,dx \tag{A10}$$

$$\tau_{yz} = \frac{Gnb}{2\pi L}\int_X^{X+L}\frac{x}{x^2 + Y^2}\,dx \tag{A11}$$

From equation (A6),

$$\tau_{xz} = -\frac{\sigma}{2}\int_X^{X+L}\frac{Y}{x^2 + Y^2}\,dx \tag{A12}$$

$$\tau_{yz} = \frac{\sigma}{2}\int_{X}^{X+L} \frac{x}{x^2 + Y^2}\,dx \tag{A13}$$

The shear stress for slip in the oblique plane is

$$\tau = -\tau_{xz} \sin 19.5^\circ + \tau_{yz} \cos 19.5^\circ \tag{A14}$$

For values of X and Y small compared with those of L, τ_{xz} is nearly independent of D:

$$\tau_{xz} \approx \sigma/6 \tag{A15}$$

$$\tau_{yz} \approx \frac{\sigma}{2} \log_e \frac{L}{D} \tag{A16}$$

At 5-percent elongation a reasonable value for L is 4×10^{-4} centimeter. If D is 400 angstroms,

$$\tau_{yz} \approx 2\sigma \tag{A17}$$

From equation (A14)

$$\tau \approx 2\sigma \tag{A18}$$

This calculation was based on the supposition that long parallel segments of the dislocation loops are in the screw orientation. This will not be strictly true. Also, equation (A6) is based on a calculation for dislocations, otherwise unrestrained, piled up against an obstacle. Actually, the segments are restrained by jogs, though to a decreasing degree going from the outer to the inner loops. The segments will not be packed so closely as indicated in equation (A6). The resolved shear stress in the oblique plane on a deflected segment will be less than that indicated in equation (A18). However, even if τ is only 50 percent of σ, it should be sufficient to push the screw segment a short distance into the oblique plane.

The repulsive force between the dislocation in the oblique plane and the dislocations in the horizontal plane will be much less if a segment nearer the center of the row of dislocations is deflected. Probably then a segment from the inside loop or the source itself, when it is in the screw orientation, is deflected. Deflection of the source means, of course, that no more loops will be generated in the plane. This is probably the principal limiting factor on the step heights of the slip lines and explains why even in a highly stressed sample very few slip lines with more than 200- to 300-angstrom step height are observed.

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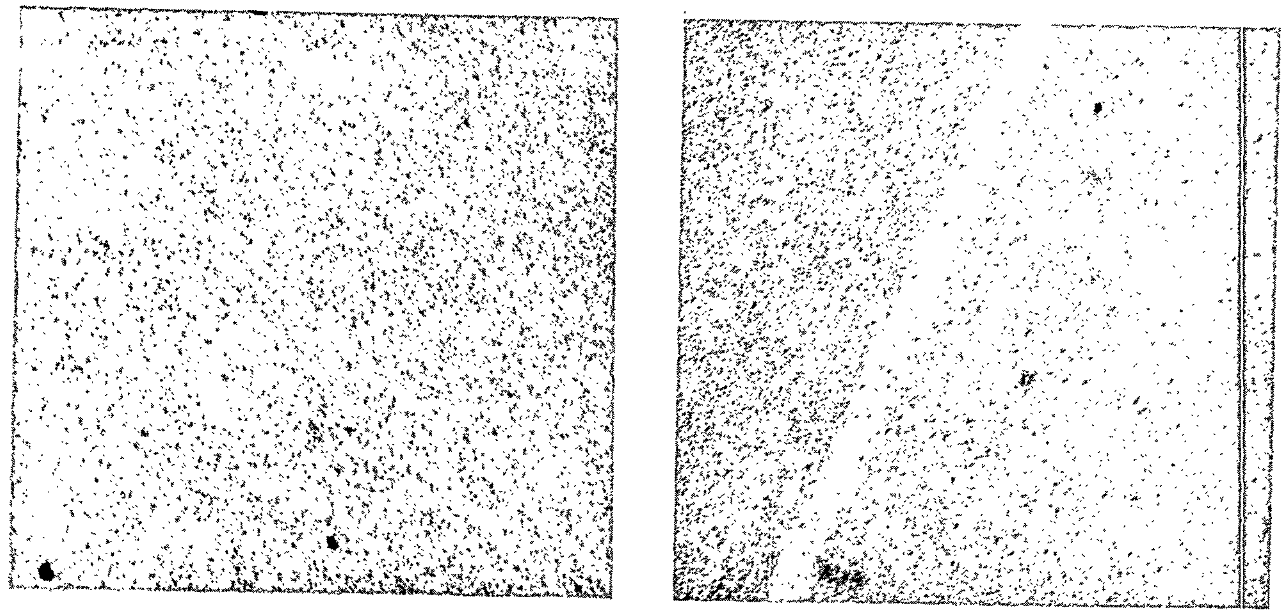
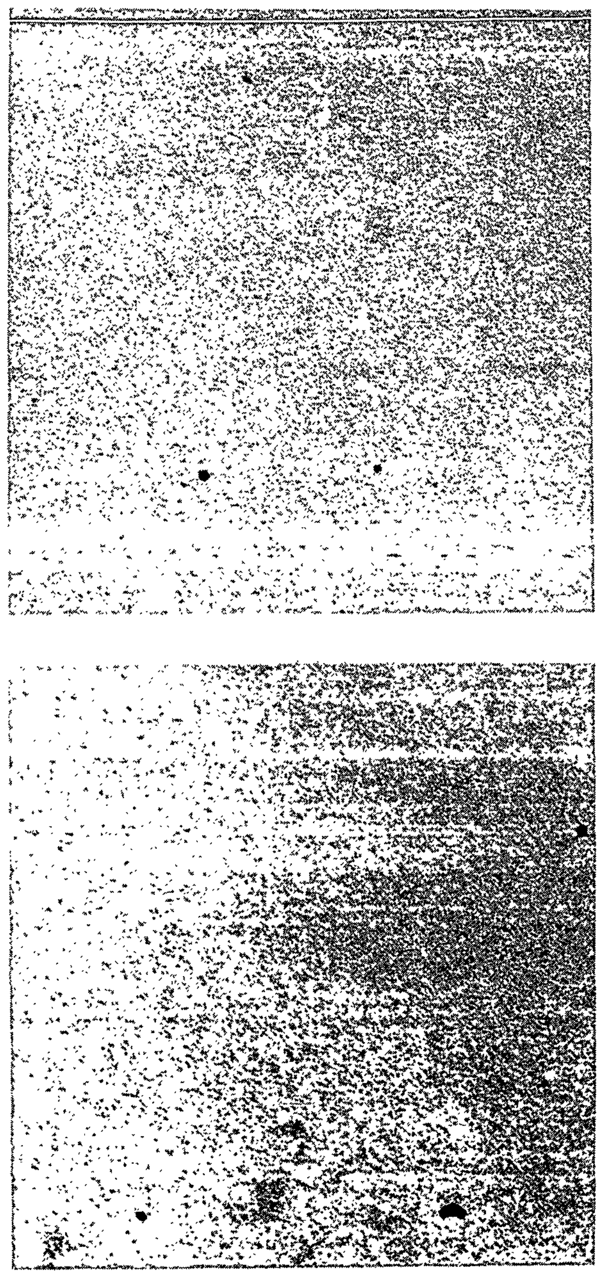


L-92457

(a) 1.5-percent elongation.   (b) 4-percent elongation.

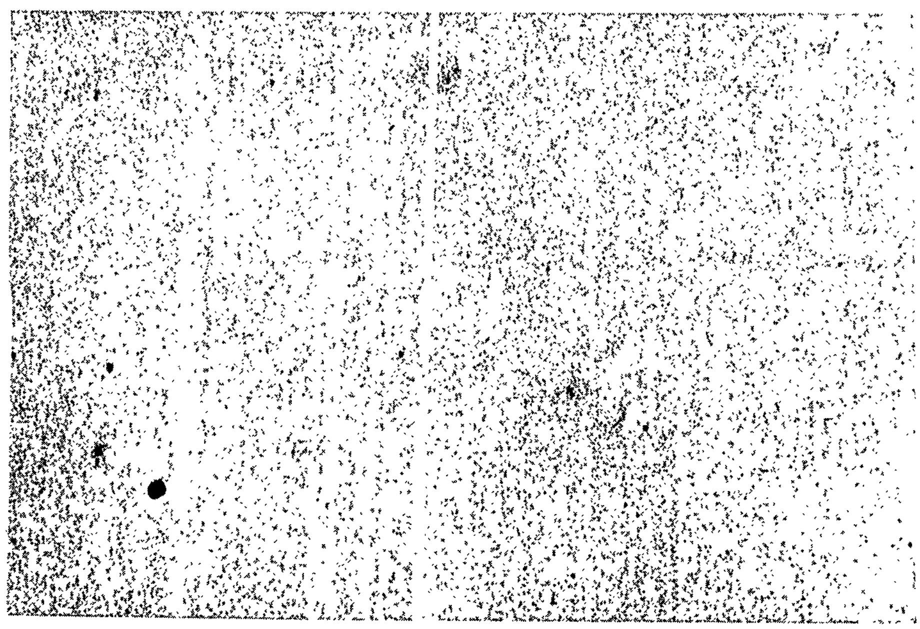
Figure 1.- Electron micrographs of slip lines in crystal 4 elongated 1.5 percent and 4 percent. 25,000X.



L-92458

(b) Concluded.

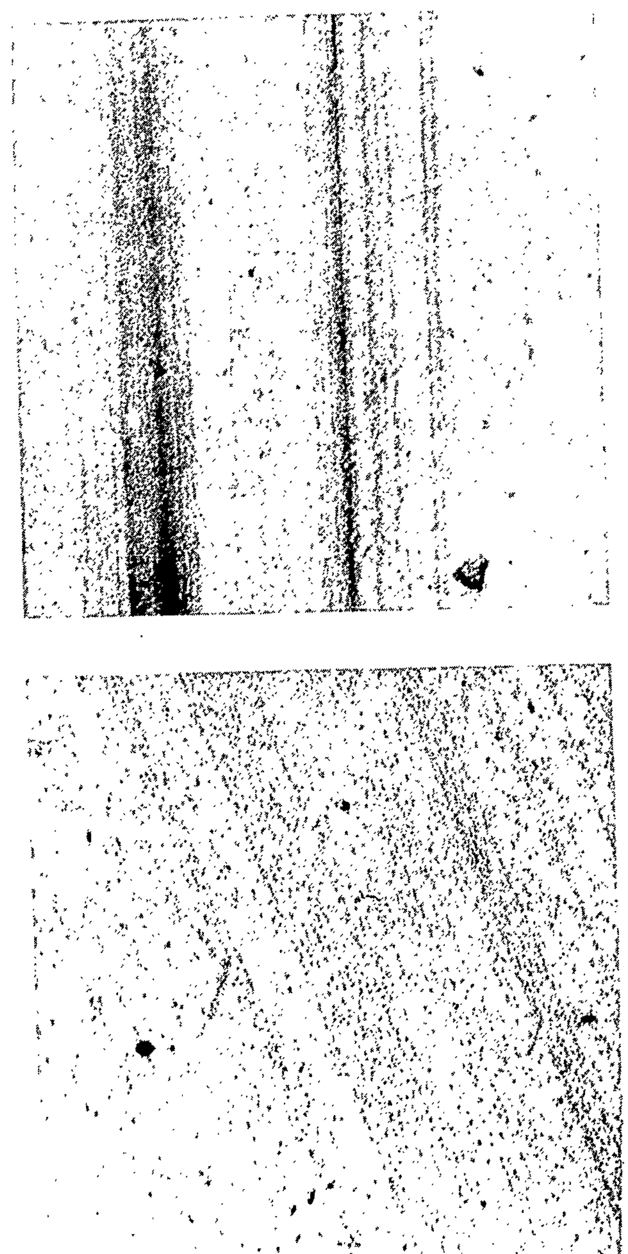
Figure 1.- Concluded.



L-92459

(a) Crystal 4; 4-percent elongation.

Figure 2.- Electron micrographs of slip lines in crystals elongated 4 percent and 10 percent. 20,000X.



L-92460

(b) Crystal S68; 10-percent elongation.

Figure 2.- Concluded.

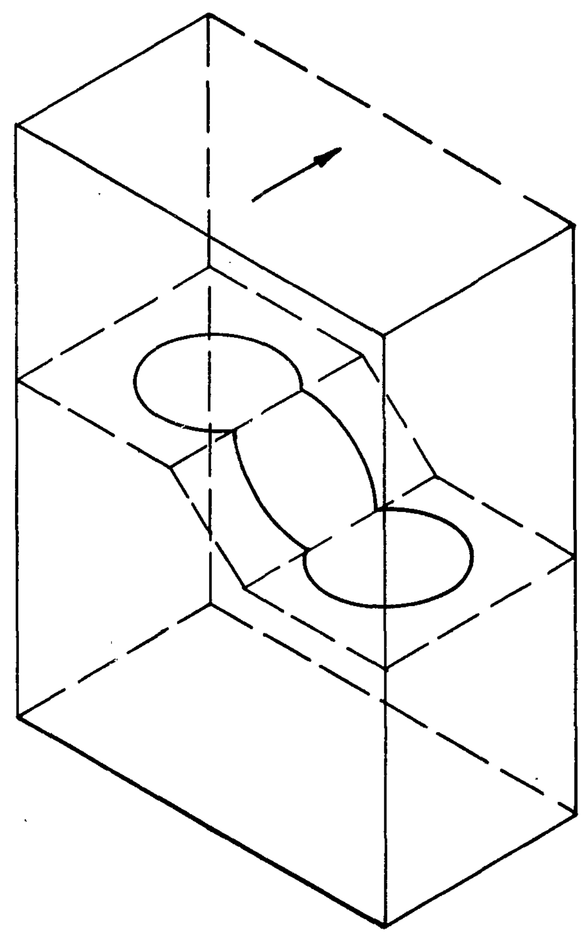
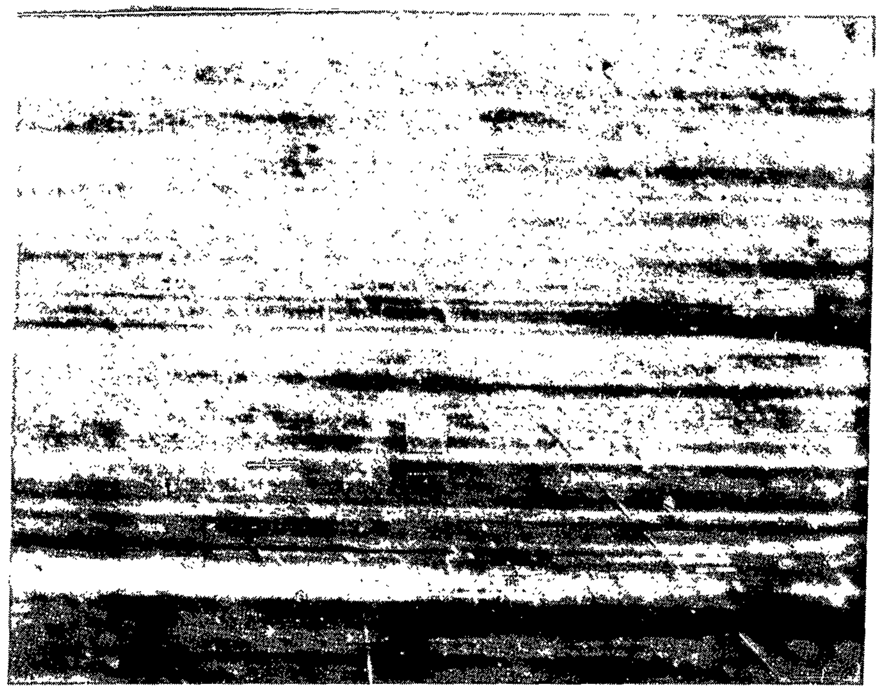


Figure 3.- Transferal of slip between primary slip planes by deflection of a screw segment in an oblique plane (from work of Orowan, ref. 9).



L-92461

(a) Crystal P68; 5-percent elongation.

Figure 4.- Light micrographs of aluminum crystals strained in tension. 1,500X.

L-92462

(b) Crystal P64; 20-percent elongation.

Figure 4.- Concluded.

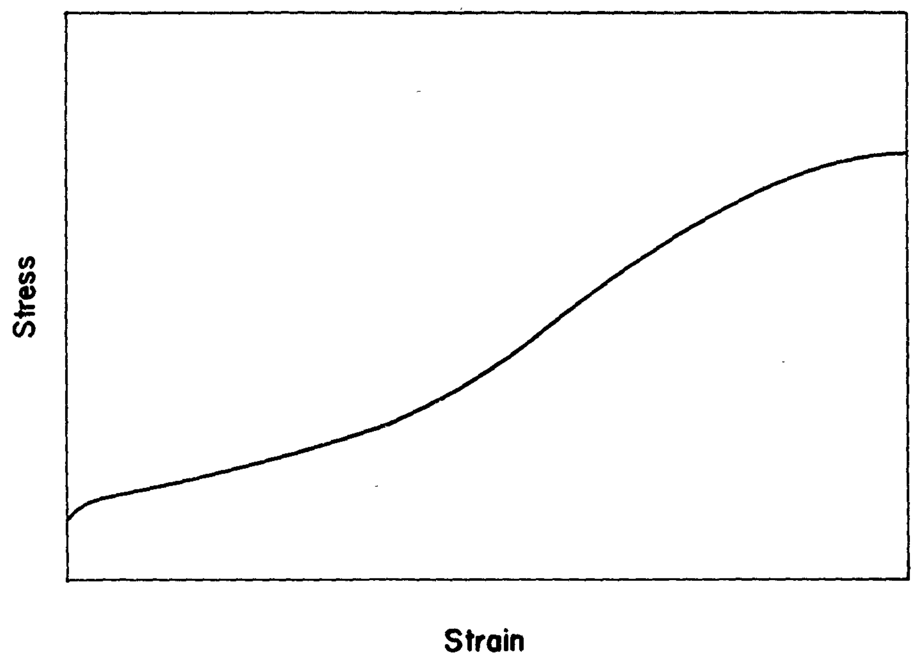
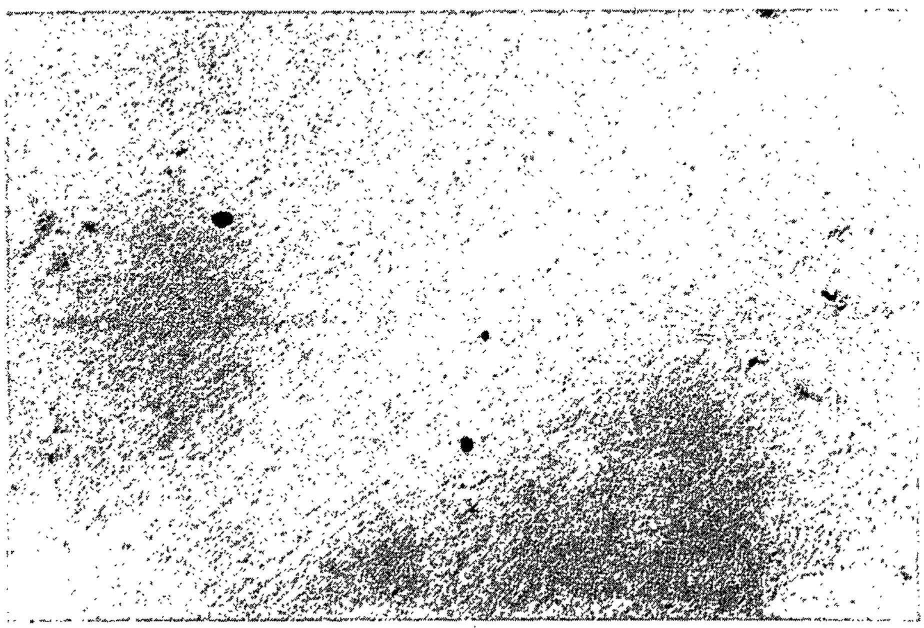


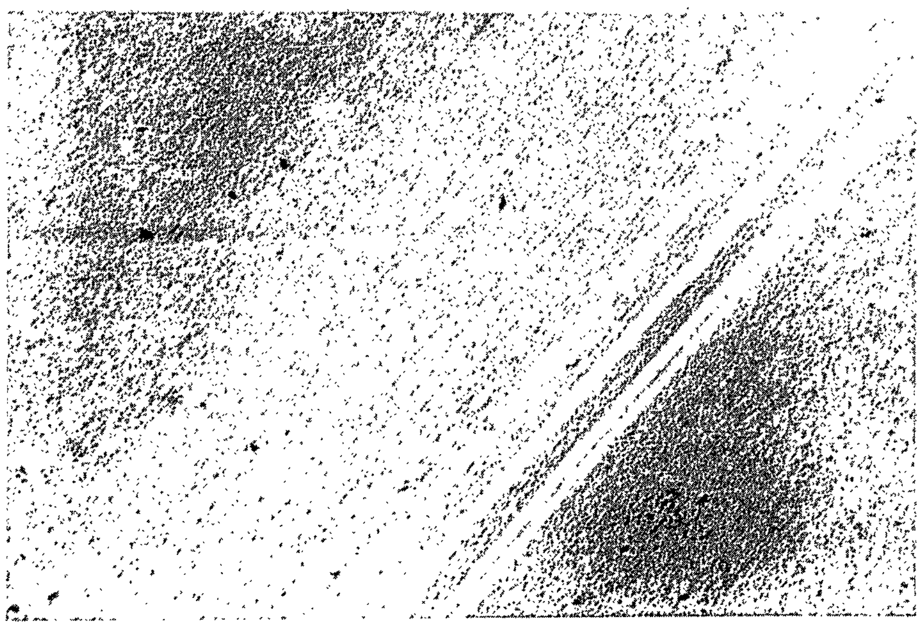
Figure 5.- Shape of stress-strain curve for an aluminum single crystal (from work of Rosi, ref. 16).



L-92463

(a) 20,000X.

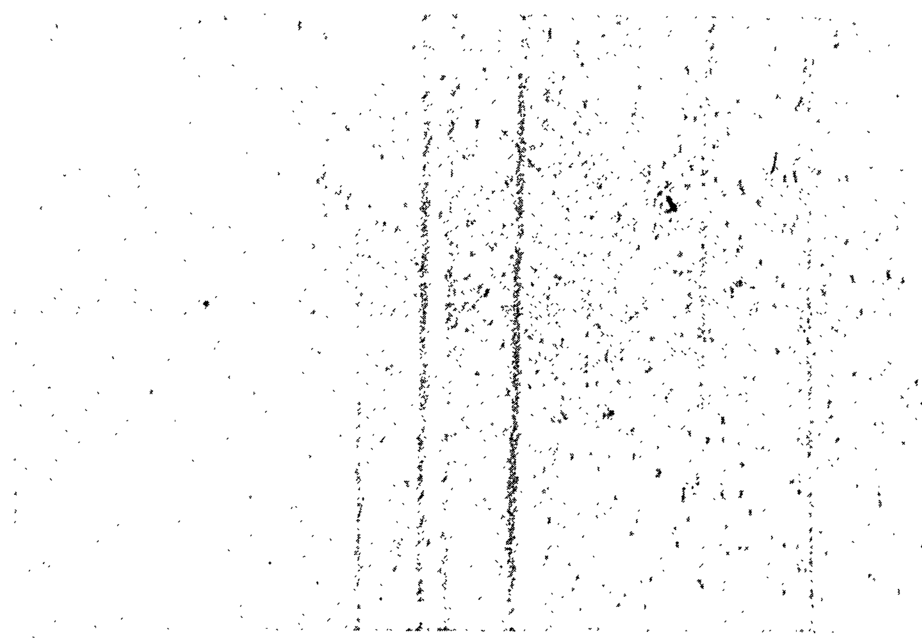
Figure 6.- Electron micrographs of slip lines in crystal 5 elongated 20 percent.



L-92464

(b) 25,000X.

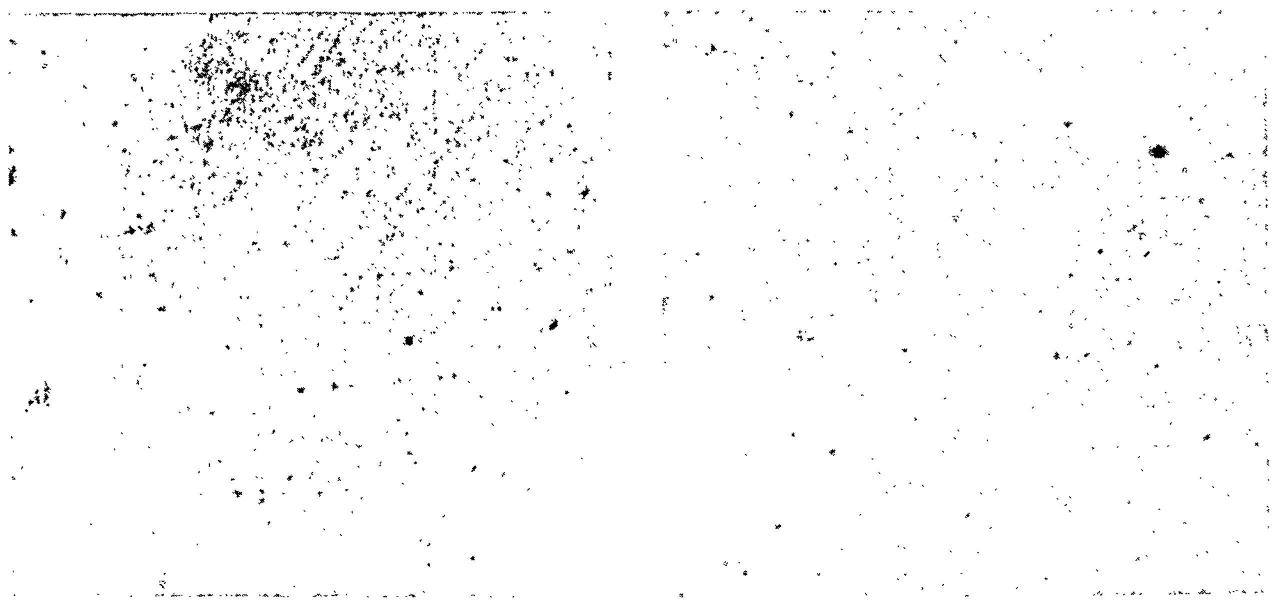
Figure 6.- Concluded.



L-92465

(a) Crystal P64; 20-percent elongation. 20,000X.

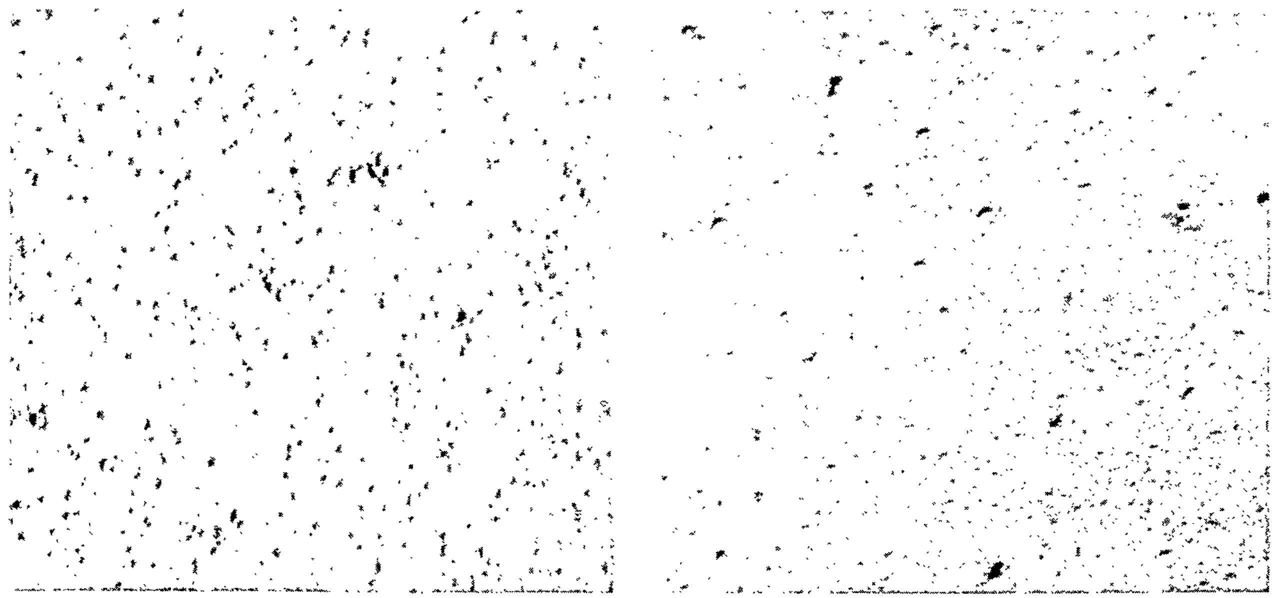
Figure 7.- Electron micrographs of slip lines in crystals elongated 20 percent and 5 percent.



L-92466

(b) Crystal 5; 5-percent elongation. 25,000X.

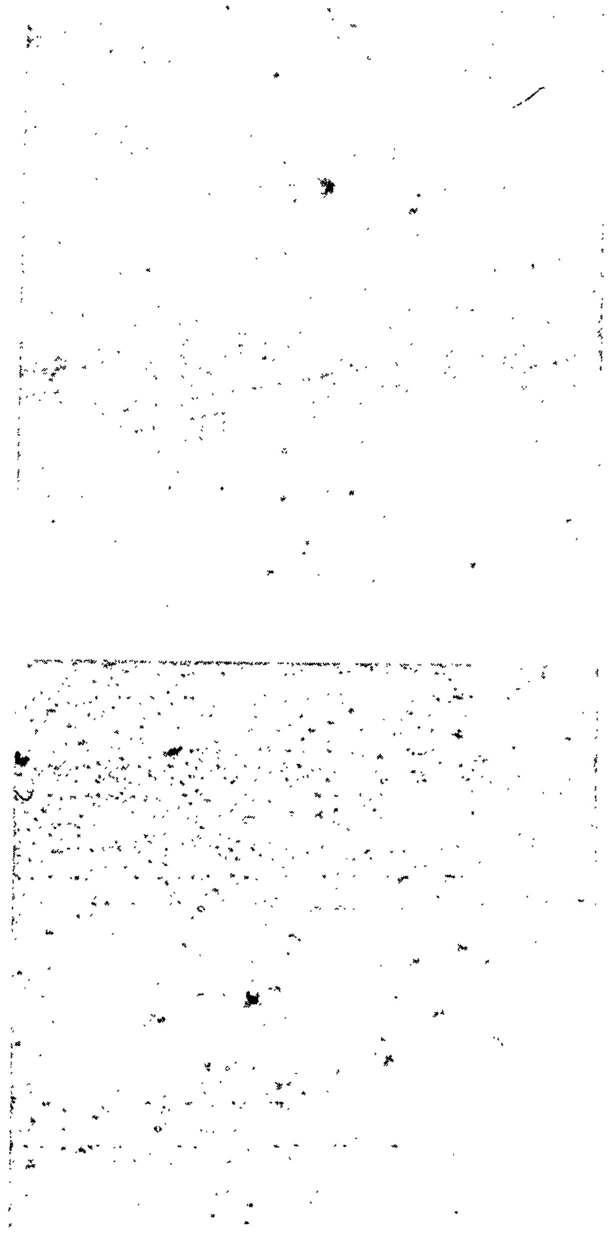
Figure 7.- Concluded.



L-92467

(a) Repolished after 5-percent elongation
and then elongated 5 percent.

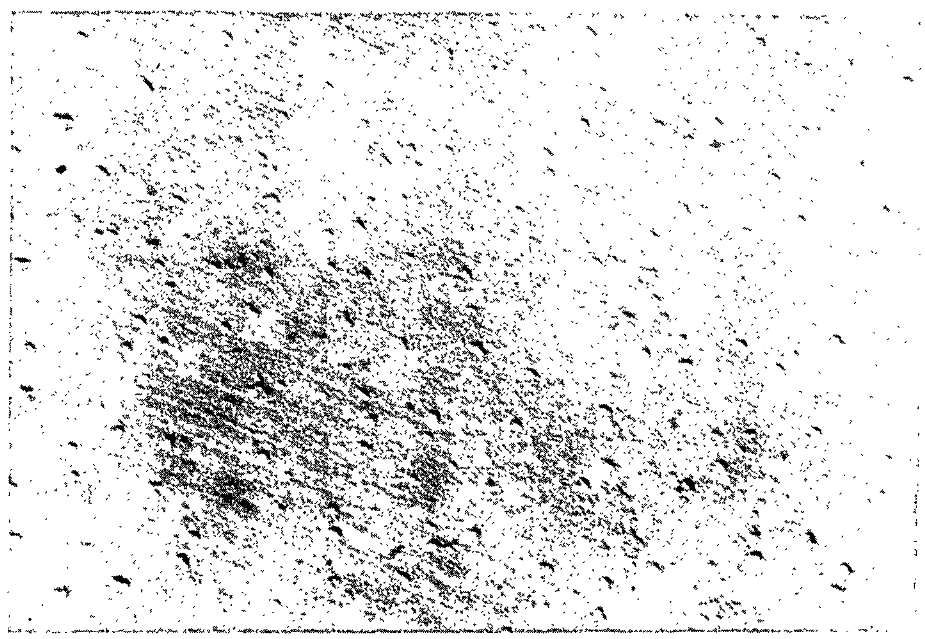
Figure 8.- Electron micrographs of two stages in slip-line formation
in crystal 5. 25,000X.



L-92468

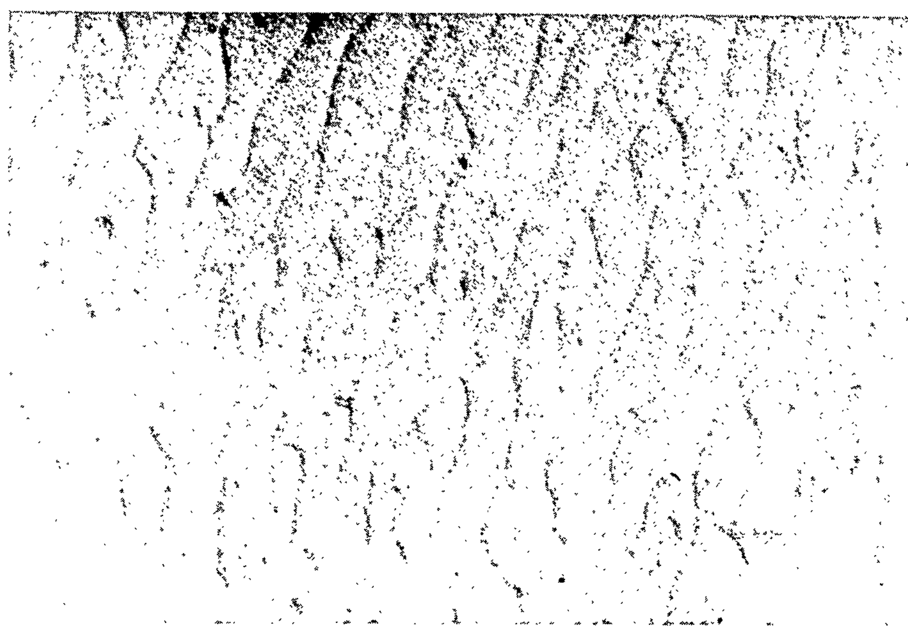
(b) Repolished after 10-percent elongation and then elongated 5 percent.

Figure 8.- Concluded.



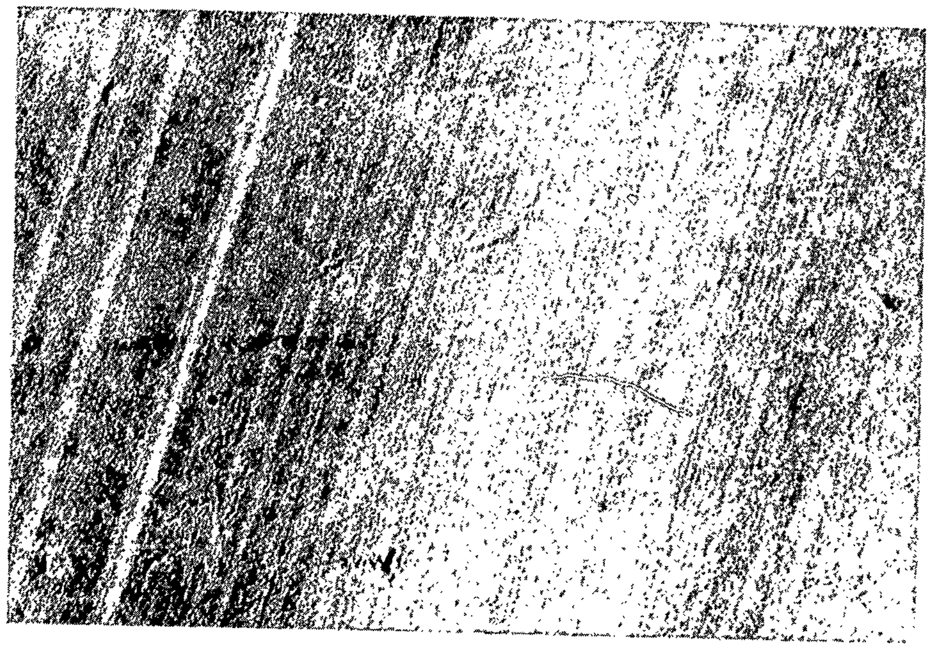
L-92469

Figure 9.- Crystal I-64 repolished after 20-percent elongation and then elongated 20 percent. 25,000X.



L-92470

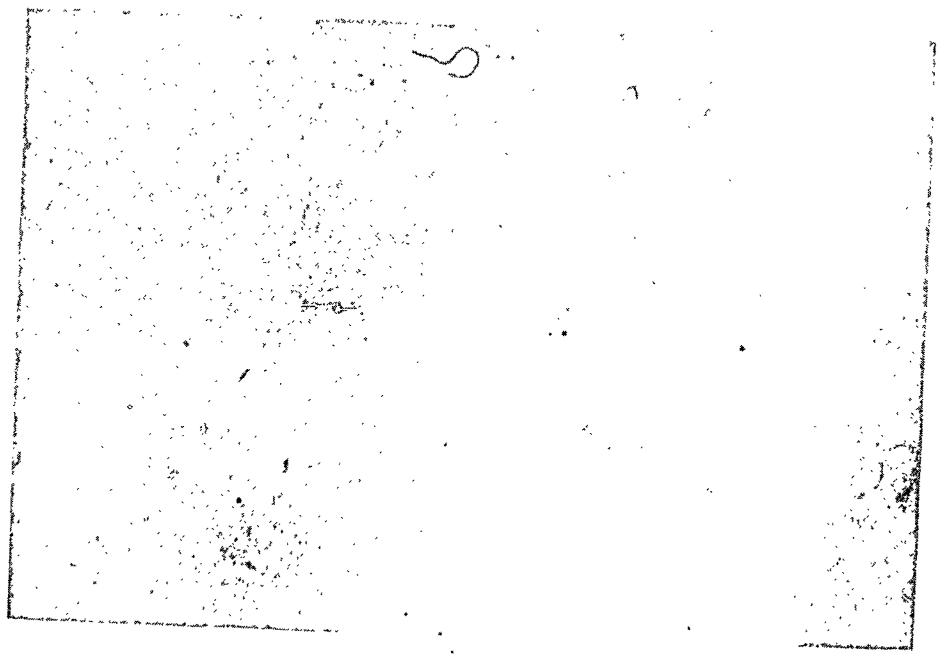
Figure 10.- Slip lines superimposed on subgrain structure in crystal S63. 20,000X.



L-92471

(a) 20,000X.

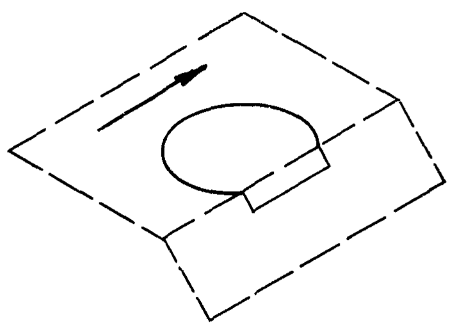
Figure 11.- Slip lines in crystal P66 elongated 10 percent at 250° F.



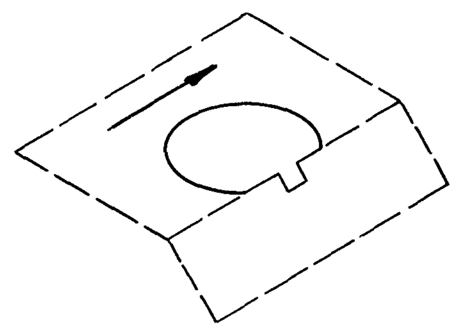
L-92472

(b) 14,000X.

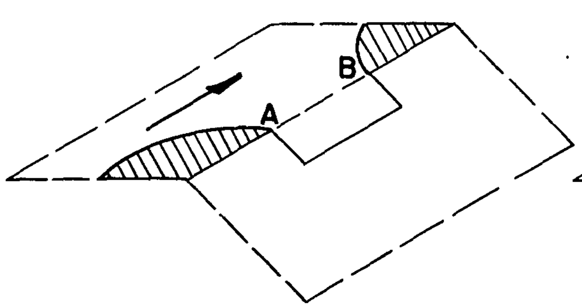
Figure 11.- Concluded.



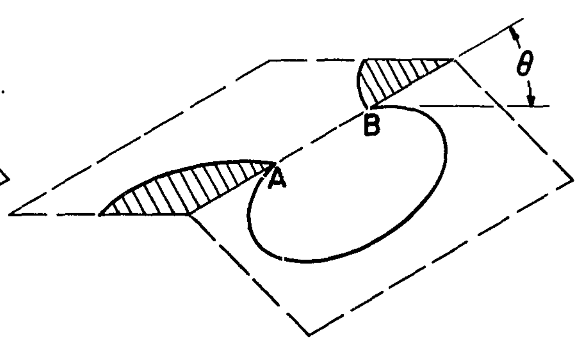
(a) Transferal by jogs of a long screw segment into an oblique plane.



(b) Transferal by jogs of a short screw segment into an oblique plane.



(c) Magnified view of segment in part (b).



(d) Dislocation loop in oblique plane after expansion.

Figure 12.- Deflection by jogs. Crosshatching represents area between partial dislocations in an extended dislocation.

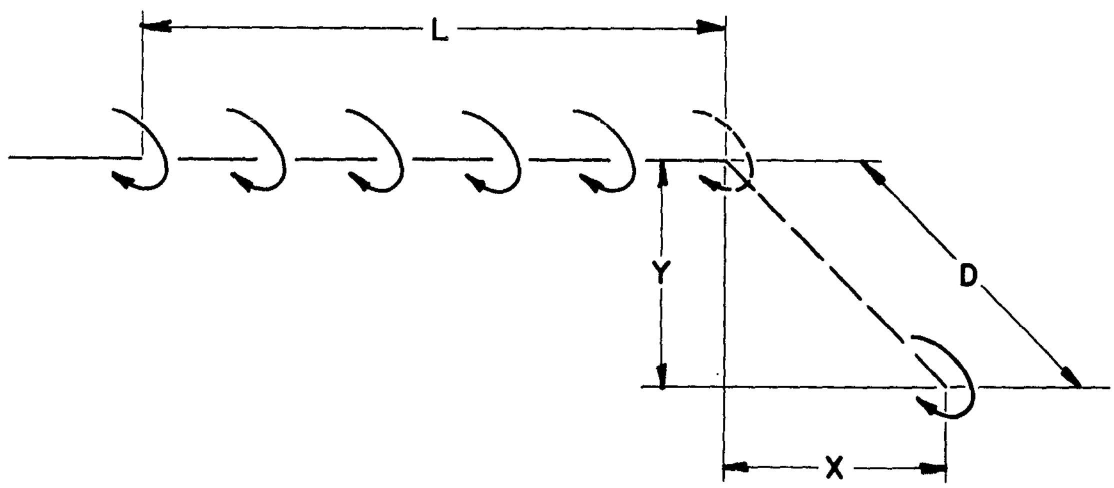


Figure 13.- End view of a row of screw dislocations in xz-plane and of a deflected screw dislocation in an oblique plane.

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STUDY OF ALUMINUM DEFORMATION BY ELECTRON MICROSCOPY. A. P. Young, C. W. Melton, and C. M. Schwartz, Battelle Memorial Institute. August 1956. 39p. diagrs., photos. (NACA TN 3728)

The slip-line structure in a series of aluminum single crystals at various stages during deformation was examined in an electron microscope. The slip-line structure was also investigated in crystals strained after various degrees of prestraining. On the basis of the data from this study and from other investigations a theory is proposed for the deformation mechanism in aluminum crystals. The possible effects of short-range ordering on deformation modes are discussed.

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